Progress towards III-V-Bismide Alloys for Near- and Mid-Infrared Laser Diodes

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Abstract—Bismuth-containing III-V alloys open-up a range of possibilities for practical applications in semiconductor lasers, photovoltaics, spintronics, photodiodes and thermoelectrics. Of particular promise for the development of semiconductor lasers is the possibility to grow GaAsBi laser structures such that the spin-orbit-splitting energy (Δ_{SO}) is greater than the bandgap (E_g) in the active region for devices operating around the telecom wavelength of 1.55 μm , thereby suppressing the dominant efficiency-limiting loss processes in such lasers, namely Auger recombination and inter-valence band absorption (IVBA). The $\Delta_{SO} > E_g$ band structure is present in GaAsBi alloys containing > 10% Bi, at which composition the alloy band gap is close to 1.55 μm on a GaAs substrate making them an attractive candidate material system for the development of highly efficient, uncooled GaAs-based lasers for telecommunications. Here we discuss progress towards this goal and present a comprehensive set of data on the properties of GaAsBi lasers including optical gain and absorption characteristics and the dominant carrier recombination processes in such systems. Finally, we briefly review the potential of GaAsBiN and InGaAsBi material systems for near- and mid-infrared photonic devices on GaAs and InP platforms, respectively.

Index Terms—Bismides, bismide alloys, GaAsBi, InGaAsBi, GaAsNbi, laser diode, optical gain, internal optical losses, absorption spectra, spontaneous emission, quantum well temperature sensitivity, Auger recombination, carrier recombination processes, efficiency

I. INTRODUCTION

THE incorporation of bismuth into III-V semiconductors such as GaAs causes a large band gap (E_g) bowing (by up to ~80 meV per % Bi), which can be modeled using band-anti-crossing (BAC) theory similarly to the dilute nitrides, albeit in the valence band (VB) for bismides versus the conduction

band (CB) for dilute nitrides [1]. As a result of this anti-crossing the VBs split into six valence sub-bands which may be grouped into the E_+ (HH^+ , LH^+ , SO^+) and E_- (HH^- , LH^- , SO^-) levels, where HH and LH refer to heavy and light holes, respectively, and SO to the spin-orbit split off band [1]. Furthermore, particularly useful properties of bismides are brought-about by the fact that due to an enhanced electron spin-orbital angular momentum interaction with the heavy Bi atom, the spin-orbit band of a bismide III-V alloy moves down in energy and coupled with the strong upward movement of the valence band edge gives rise to a significant increase in the spin-orbit splitting energy (Δ_{SO}) [2], which enables the suppression of Auger losses in lasers [3, 4] and may be exploited in the field of spintronics [5]. Figure 1 presents a summary of experimental data of the compositional dependence of E_g and Δ_{SO} in GaAsBi as a function of Bi concentration. From these data it is predicted that the energy cross-over $\Delta_{SO} = E_g$ occurs at bismuth concentrations around 10%, which is in good agreement with theory for free-standing GaAsBi [2, 6].

The practical realisation of the condition $\Delta_{SO} > E_g$ promises to provide a bandstructure leading to the suppression of the main efficiency-limiting loss processes in infrared lasers, namely Auger recombination, involving a Conduction band electron, Heavy hole and Spin split-off hole excited to Heavy hole band (“CHSH”) [7], and inter-valence band absorption (IVBA). In the IVBA process, emitted photons are re-absorbed in the active region, thereby increasing the internal optical losses and negatively impacting upon the characteristics of infrared emitters such as conventional 1.55 μm InGaAsP/InP lasers [3, 4, 8, 9]. 60-80% of injected current in these devices is dissipated at threshold as heat due to CHSH Auger recombination [3, 8]. In combination with IVBA, degrades both the device external differential efficiency and threshold current, these processes make laser characteristics very sensitive to the operating temperature requiring additional temperature control. The development of uncooled laser modules, would result in a significant energy savings. At the moment, more than 95% of energy consumed by a commercial telecom laser diode is used by a thermoelectric cooler) and hence removing or reducing these processes is highly attractive. Providing uncooled operation

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would allow energy savings in excess of 50% of current values. Thus, at least 50% of the power demand in a fibre-to-the-home (FTTH) application could potentially be saved [10].

There have been a number of previous attempts to improve the thermal performance of existing 1.55 μm lasers. The InGaAlAs/InP material system has been studied extensively owing to the fact that its increased conduction band offset helps to reduce electron overflow. However, it has only a relatively small effect on Auger recombination [3, 8]. The GaAs-based dilute nitride material system is attractive, but devices tend to suffer from large defect-related recombination in addition to Auger recombination which continues to play a major role [11]. 1.3 μm InAs/GaAs [12] and 1.5 μm InAs/InP [13] quantum dot based lasers have similarly been shown to suffer from Auger recombination and have comparable temperature sensitivity to conventional quantum well lasers unless p-doping is employed which also causes an increase in the threshold current [14]. Achieving infrared lasers with both low threshold currents and low temperature sensitivity therefore remains a challenge owing to the fundamental role of Auger recombination.

Fig. 1. The experimental Bi composition dependence of the room temperature spin-orbit splitting energy (Δ_{SO}) and unstrained bandgap energy (E_{g0}) for free-standing GaAsBi. The dashed curves are guides to the eye. The symbols show the results of different authors [2, 5, 15, 19, 20, 21, 25].

In 2009, Sweeney proposed that Bismuth-containing alloys offer a potential way forward since both the strong bandgap reduction and increasing Δ_{SO} that arise with the addition of Bi open-up the possibility of CHSH Auger and IVBA suppression when $\Delta_{SO} > E_g$ [9, 16]. This promises efficient and thermally stable photonic devices in the datacom/telecom range of 1.3–1.6 μm . This also provides a route to devices such as efficient 1.55 μm monolithic vertical cavity surface emitting lasers (VCSELs) and related devices grown on GaAs [17]. Since this time there has been significant effort into the development of bismide alloys reflected through a strong increase in publications on this topic [18].

A thorough understanding of the fundamental physical properties of bismuth-containing alloys is therefore very important for the development of this system. Whilst there have been reports in the past 10–15 years on the structural,

optical [5, 19, 20], thermal [21, 22] and transport properties [23] of bismides, they remain a relatively under-explored family of III-V alloys. Bismides have also attracted interest owing to their potential to have a reduced temperature sensitivity of the band gap, dE_g/dT , compared to conventional III-Vs due to the fact that InBi and GaBi are semi-metallic compounds [24]. Some reports indeed indicate dE_g/dT is significantly weaker in GaAsBi [21], but there is uncertainty [25, 26] regarding the extent to which this is an intrinsic property or related to material inhomogeneity. There is relatively little information on dE_g/dT for the quaternary alloy, InGaAsBi; Devenson et al. reported a qualitatively weaker dE_g/dT compared to InGaAs [27], however, Marko *et al.* report relatively little influence of Bi content on dE_g/dT [28]. This is crucial to understand not only in terms of the underlying physics, but also for the extent to which the laser emission wavelength may be stabilised with temperature through the use of these alloys. Together with the suppression of non-radiative recombination, a temperature insensitive emission wavelength could significantly improve the temperature stability of lasers.

Since bismuth principally influences the valence band, while nitrogen influences the conduction band, combining bismuth and nitrogen in III-V alloys offers huge potential for engineering the conduction and valence band offsets, the strain, bandgap and spin-orbit splitting, with wide scope for the design of photonic devices, as discussed in Section VI of this paper [17].

GaAsBi/GaAs is suitable for near-IR, particularly 1.5 μm telecommunications devices; InGaAsBi [29] or GaAsBiN/GaAs [17] have the potential to span a wide spectral range from the near-infrared out into the mid-infrared ($<6 \mu\text{m}$), while still maintaining tolerable strain levels on standard InP or GaAs substrates. The growth and processing of the current mid-infrared GaSb-based materials is less advanced than for InP-based materials, which are being widely used commercially for optical fibre based telecommunication applications. Growth on the InP material system benefits from well-known fabrication techniques and the availability of low-loss waveguides. The use of standard GaAs substrates for near- or mid-infrared applications is also beneficial from the point of view of well-established technology and possibility for the growth of lattice-matched Distributed Bragg Reflectors (DBRs) for VCSELs and related devices.

The main part of this paper discusses the properties of GaAsBi based lasers on GaAs. In section II, we review the main approaches in bismide laser development and, in section III, we discuss experimental results on optical gain and absorption properties. Then, in sections IV and V, characteristics and achievements in the development of GaAsBi laser diodes are discussed. In section VI, new possibilities of GaAsBiN/GaAs and InGaAsBi/InP material systems are presented. Finally, in section VII, we summarize main achievements and difficulties in the development of bismide-based photonic devices.

II. THE DEVELOPMENT OF GAASBI LASERS

As evident from Fig. 1, the main challenge to achieve the $A_{SO} > E_g$ condition on GaAs is to incorporate a relatively high Bi composition (>10%) whilst maintaining the high material quality as required for laser structures. Due to the metastable nature of GaAsBi a much lower growth temperature is required compared to the well-established growth of (Al)GaAs alloys. This presents a significant challenge in terms of avoiding the formation of defects. Various approaches have been used to fabricate GaAsBi based laser diodes as summarised in Fig. 2. The first GaAsBi based laser was realised under optical excitation and consisted of a bulk-like 390 nm thick $\text{GaAs}_{0.975}\text{Bi}_{0.025}$ active layer grown using molecular beam epitaxy (MBE), which was reported by Tominaga *et al.* [30]. Optically pumped pulsed operation was achieved up to a temperature of 240 K. This was followed by Ludewig *et al.* who demonstrated the first electrically pumped GaAsBi laser with a Quantum Well (QW)-based active region [31]. This device, which was grown by metal-organic vapour phase epitaxy (MOVPE), consisted of a $\text{GaAs}_{0.978}\text{Bi}_{0.022}$ single QW active region within an $\text{Al}_{0.2}\text{Ga}_{0.8}\text{As}$ waveguide and $\text{Al}_{0.4}\text{Ga}_{0.6}\text{As}$ cladding layers on an n-doped GaAs (001) substrate. The GaAsBi QW of the devices with 2.2% Bi was grown at 400°C under pulsed precursor flow [31]. Electrically pumped pulsed operation was demonstrated at room temperature with a threshold current density of 1.56 kAcm^{-2} [31]. More recently, Kim *et al.* demonstrated an MOVPE grown GaAsBi SQW laser with 2.5% Bi operating at room temperature with a threshold current density of 4.1 kAcm^{-2} and an emission wavelength of 960nm. Their structure utilised $\text{GaAs}_{0.8}\text{P}_{0.2}$ barriers which provide a beneficial increase in the band offsets while compensating for the QW strain [32, 33].

In order to increase the Bi fraction a continuous precursor flow needs to be applied that enables higher growth rates under conditions of the optimised V/III and trimethyl bismuth/V ratios at 400°C growth temperature [34]. Using this approach, the Bi composition in GaAsBi QW lasers was increased up to 4.4% in MOVPE grown structures. Lasing was observed in these devices up to 180 K [35]. It becomes increasingly difficult to achieve higher Bi compositions due to the need to maintain sufficiently high growth temperatures (>350°C) to decompose the metal organic precursors [34]. The other issue related to MOVPE growth at lower temperatures is caused by hydrocarbon rest groups from the precursor decomposition, which stick to the surface, cannot desorb and in turn alter the growth conditions in ways, which are not observed for high temperature growth. There are also problems with the formation of metallic Bi films riding the surface and with droplets since surplus Bi segregates to the surface and does not desorb at the relatively low temperature [36]. Therefore, there is an influence of group V gas phase conditions on the growth rate, which is in general not observed for high-temperature growth, but which makes control of the growth conditions challenging. A comprehensive review of the MOVPE growth mechanisms of bismide III-V alloys can be found in [37].

Higher Bi fractions (>10%) have been achieved by MBE

growth [38, 39] where the growth temperature can be reduced to as low as 200°C. Bismuth fractions as high as 22 % have been reported by MBE [40], although these samples suffered from metallic droplets on the surface. The bandgap and optical absorption edge study of MBE grown $\text{GaAs}_{1-x}\text{Bi}_x$ alloys with $0 < x < 17.8\%$ has also been recently reported [41]. However, to grow high-quality (Al)GaAs barriers and waveguide/cladding layers the growth temperature should be significantly increased, without negatively affecting the composition and material quality of the active region. A comprehensive review of MBE growth of GaAsBi can be found elsewhere [42].

The longest wavelength of lasing operation at room temperature was observed for a $\lambda=1204 \text{ nm}$ device under optical excitation in $\text{GaAs}_{0.941}\text{Bi}_{0.059}$ bulk MBE layers with thicknesses of 300-400 nm [43]. Interestingly, all the samples independently of Bi% in [43] showed the same characteristic temperature, T_0 ($T_0 = [\text{dln}(J_{th})/\text{dT}]^{-1}$ with J_{th} being the threshold current density at device temperature T) of the lasing threshold of 100K in the temperature range of 20-80°C. This value is also the same value as previously reported in 2.2%Bi QW lasers above room temperature and may be attributed to the dominant defect related non-radiative recombination [35]. Recently, MBE grown GaAsBi bulk diode lasers have been demonstrated with up to 4% Bi emitting at room temperature at a wavelength of 1045 nm. Being bulk lasers, their threshold current density is correspondingly high at $8\text{-}15 \text{ kA/cm}^2$ [44] (see Fig. 2).

Fig. 2. Threshold current density for GaAsBi based lasers [33, 34, 35, 43, 44, 46] as a function of Bi composition. Arrows indicate Bi composition in the longest wavelength GaAsBi bulk optically pumped lasers (typical lasing threshold of 0.36 mJ/cm^2 at 300 K [43]).

To overcome the challenges associated with MBE and MOVPE growth, a hybrid approach has been developed [45, 46]. This hybrid growth method consists of growing the QW region using MBE (thereby allowing an increased Bi composition to be achieved in the active region), while the remainder of the laser structure is grown using MOVPE thereby maintaining the capability for rapid high quality

growth of the thick waveguide and cladding layers. This method allows for an increase of Bi composition in the active region above 6% but still maintains the advantages of the MOVPE and MBE approaches. Using the hybrid growth method, GaAsBi/GaAs lasers consisting of three QWs containing ~6% Bi were developed, and electrically pumped operation was demonstrated at room temperature [46]. Laser structures with as high as 8% Bi in QWs were produced, but have not yet demonstrated lasing operation [46]. Characterisation of these devices using temperature dependent measurements of stimulated and pure spontaneous emission as well as high hydrostatic pressure has revealed that performance is limited by recombination via defect states and significant inhomogeneous broadening of the carrier distribution due to compositional inhomogeneity of the active region [35, 46]. Structures with high Bi fraction exhibited issues related to poor carrier transport in QWs and recombination in the barriers [46]. The detailed theoretical investigations of the emission dynamics [47] and optical spectra in GaAsBi alloys highlighted the strong role played by localised states associated with Bi clustering in determining the optical properties of GaAsBi epitaxial layers. One can see from Fig. 2 how the device performance deteriorates with increasing Bi fraction. A relatively high threshold current density of bismide laser diodes with ~2% Bi becomes even higher with increasing Bi composition in the active region, which demonstrates that further improvement and optimisation of the growth and fabrication of GaAsBi materials and devices is required in order to realise their potential for practical applications.

In addition to refinement of growth and fabrication processes, an important aspect of the development of GaAsBi material system is to develop a quantitative understanding of the impact of Bi incorporation on the properties and performance of existing devices to enable the design and optimisation of longer-wavelength devices with increased Bi composition. Such optimisation requires a detailed understanding of the physical properties of the laser structures to develop improved device designs with optimised QW thickness, alloy composition, strain and band offsets, as well as QW number and the waveguide refractive index profile, etc. to deliver the required laser characteristics such as high optical gain, output power and efficiency, and temperature stability.

III. OPTICAL GAIN IN GAASBI/GAAS LASER STRUCTURES

There have been only a few initial theoretical investigations of optical gain owing to a lack of detailed information regarding the band structure of GaAsBi alloys. These limitations were overcome by undertaking a series of detailed analyses of the electronic properties of GaAsBi alloys [48, 49]. Using this improved understanding of the GaAsBi band structure a most comprehensive theoretical model for dilute bismide quantum well lasers was developed and applied to elucidate the effects of Bi incorporation on the electronic and optical properties of ideal GaAs-based laser structures operating at wavelengths up to 1.55 μm [50].

This theoretical model was applied to compute, analyse and

compare to experiment the spontaneous emission (SE) and optical gain spectra of the GaAsBi SQW laser structures. We obtained the first experimental measurements of the optical gain spectra of GaAsBi/(Al)GaAs QW lasers [51] by applying the segmented electrical contact method [52] allowing direct measurements of the absorption, gain and spontaneous emission (SE) spectra as well as of the internal (cavity) optical losses in GaAsBi laser structures containing ~2% Bi [51].

By performing such measurements from multi-sectional devices at different current densities the net modal gain spectra were obtained. The net modal gain is determined as the modal gain G (the product of material gain, g , and the optical confinement factor, Γ) minus internal optical losses, α_i , and is presented in Fig. 3. The experimental data plotted with different symbols were measured at different current densities of 0.7, 1.4, 2.0 and 2.4 kA/cm^2 [51]. As can be seen from Fig. 3, the measured gain spectra are relatively broad, with a full width at half maximum of ~100 meV at a current density of 2 kA/cm^2 , which is close to twice that observed for an InGaAs/GaAs SQW laser operating at a similar wavelength [53]. The reason for this is related to the strong Bi-induced inhomogeneous broadening that is evident in the optical spectra of GaAsBi alloys [35, 54]. From the value of the net modal gain at longer wavelengths ($G=0$), the value of the internal optical losses can be obtained. From Fig. 3 (and also from direct absorption spectra measurements) [51], the value of optical losses of $\alpha_i = 15 \text{ cm}^{-1}$ was determined. This is relatively high for a device at this wavelength which may indicate additional optical scattering losses associated with the material inhomogeneity.

Fig. 3. Net modal gain spectra for the same device measured using the segmented contact method (open symbols) and calculated (solid lines) at injected current densities of $J = 0.7, 1.4, 2.0$ and 2.4 kA/cm^2 corresponding to carrier densities in the theoretical calculations $n = 5.12, 7.11, 8.24$ and $9.38 \times 10^{18} \text{ cm}^{-3}$, respectively [51].

IVBA is expected not to be significant at such short wavelengths. This gives rise to an estimated peak modal gain at the current density of $J = 2 \text{ kA/cm}^2$, which is close to the threshold current density in the Fabry-Perot lasers made from

this material (see Fig.2 and also [34, 35], of $G_{\text{peak}} = 24 \text{ cm}^{-1}$. Using the calculated optical confinement factor of 1.60% for this $\text{GaAs}_{0.982}\text{Bi}_{0.018}$ SQW laser structure [51], a peak material gain of $g_{\text{peak}} \approx 1500 \text{ cm}^{-1}$ at $J = 2 \text{ kA/cm}^2$ was estimated, which agrees well with the calculated value of 1560 cm^{-1} [51] and comparable with the optical gain values in InGaAs-based structures [55].

The calculated modal gain spectra are also shown in Fig. 3 and demonstrate very good agreement with experiment. For such comparison, the carrier density corresponding to each current density, at which the gain spectra were measured, was determined as described in detail elsewhere [51]. Following that procedure, it was found that the current densities of 0.7, 1.4, 2.0 and 2.4 kA/cm^2 , at which the gain spectra were measured, occur for quasi-Fermi level separations of 1.375, 1.409, 1.423 and 1.435 eV , which in turn correspond to carrier densities in the theoretical calculations of $5.12, 7.11, 8.24$ and $9.38 \times 10^{18} \text{ cm}^{-3}$, respectively. It should be noted that the theoretical gain spectra for the current densities $0.7\text{-}2 \text{ kA/cm}^2$ include E1-HH1 optical transitions only, while the spectrum at the highest current density of 2.4 kA/cm^2 includes also E1-LH1 optical transitions. Without these transitions at the highest carrier density the peak modal gain would be underestimated by $\sim 10\%$, highlighting that TE-polarised optical recombination involving light-hole-like states plays a role at higher levels of injection. Overall, one can see from Fig. 3 that the theoretical spectra are in good quantitative agreement with the experimental data. The calculated magnitude of the net modal gain is in excellent agreement with that measured using the segmented contact method across the full investigated range of current densities, confirming the accuracy of the optical transition matrix elements derived within the framework of the 12-band **k.p** band structure model [50], and the overall shape of the experimental gain spectrum is well reproduced at each current/carrier density by the theoretical model.

From measurements of the gain spectra for several devices, fabricated from the same wafer it was possible to determine the variation of the peak modal gain G_{peak} (closed symbols) as a function of current density, which is presented in Fig. 4. The threshold modal gain G_{th} determined as the sum of the cavity (α_i) and mirror (α_m) losses, $G_{\text{th}} = \alpha_i + \alpha_m$, which is equal to the peak modal gain at threshold is also shown in Fig. 4 with open symbols. As these Fabry-Perot devices were fabricated from the same part of the wafer as the devices for which the gain spectra of Fig. 3 were measured, it was assumed that they have the same optical losses of $\alpha_i = 15 \text{ cm}^{-1}$ and the facet (mirror) losses were calculated in each case as $\alpha_m = (1/L_c)\ln(1/R)$, where R is the facet reflectivity [51]. Returning to Fig. 4, we note that the peak modal gain data obtained in this manner for the Fabry-Perot devices fits well with the overall trend observed for the multi-section devices upon which the segmented contact measurements were performed.

Based on the correspondence between the experimental current and theoretical carrier densities determined in the analysis of the gain measurements shown in Fig. 3, the variation of G_{peak} with current density was calculated. These

results are depicted by the solid line in Fig. 4. It can be seen that the theoretical calculation is again in good quantitative agreement with the measurements, confirming that the theoretical model is capable of describing the GaAsBi gain spectra across a wide range of current densities, as well as for a variety of multi-section and Fabry-Perot devices.

The experimental results on optical gain coupled with theoretical model accounting for the measured inhomogeneous broadening [51] provide the most detailed insight to date into the optical properties of QW lasers based on this novel material system. The developed model and calculations are in good quantitative agreement with the experiment, confirming our theoretical understanding of the unusual electronic and optical properties of GaAsBi laser structures and verifying the predictive capacity of this theoretical model for use in the design and optimisation of dilute bismide semiconductor lasers.

Fig. 4. Measured (various symbols) and calculated (solid line) variation of the peak modal gain as a function of injected current density for a range of multi-section (closed symbols) and Fabry-Perot (open circles) devices [51]. The threshold gain data for the Fabry-Perot devices were determined as the sum of the cavity and mirror losses at threshold current density [51].

IV. GAASBI/GAAS LASER CHARACTERIZATION AND OPTIMIZATION

In this section we consider effect of Bi composition on laser device performance using results on electrically pumped GaAsBi/(Al)GaAs QW lasers grown by MOVPE with different Bi compositions in the active region. Single quantum well (SQW) and triple quantum well (TQW) broad area laser structures were formed using $50 \mu\text{m}$ and $100 \mu\text{m}$ wide Au/Cr metal stripes deposited on the top contact and an Au/AuGe/Cr-based contact was deposited on the substrate [31, 35]. To reduce lateral current spreading the top p^+ GaAs:Zn-contact layer was etched-off using the metal stripes as a mask. Since the devices were grown on GaAs-substrates the laser facets were cleaved using standard techniques with a Fabry-Perot cavity length of 1 mm and measured *as-cleaved*. The composition of the material, the growth rates and the material uniformity were assessed using high resolution X-ray diffraction, photoluminescence and transmission electron microscopy using benchmarking samples prior to the growth of the laser structures and showed very good material quality and structural uniformity of the active region [31, 36].

The GaAsBi-based laser diodes were characterised using

electrical and optical measurements as a function of temperature using a closed cycle cryostat (20-300 K). Lasing output and emission spectra from the facet as well as pure spontaneous emission collected from a 100 μm diameter window milled in the substrate contact [56] were studied as a function of temperature. In addition, high hydrostatic pressure measurements using a He-gas pressure system were used to investigate recombination and loss mechanisms in the lasers [56, 57]. Pulsed measurements with pulse widths of 200-500 ns and repetition frequencies of 1-10 kHz were used to avoid self-heating effects. Continuous wave mode was also used for voltage-current characteristics for currents below 50 mA ($J < 100 \text{ Acm}^{-2}$).

For device modelling, the calculations of E_g and Δ_{SO} were based on recent experimental and theoretical studies of GaAsBi on GaAs using the valence band anti-crossing model (VBAC) including the effects of strain [35, 58]. The band alignment and the energy states of electrons, heavy and light holes of GaAsBi/(Al)GaAs quantum well structures were determined using these parameters with the nextnano software package, described in more detail elsewhere [17, 35].

As mentioned above, the reduction in band gap due to the addition of Bi to GaAs is mainly due to the upward movement of the valence band due to band anti-crossing. When grown on GaAs, this therefore gives rise to a large valence band offset. The conduction band moves downward to a smaller extent consistent with conventional alloying leading to a smaller conduction band offset. Hence, to improve electron confinement Al_{0.2}Ga_{0.8}As barriers were used in samples with low Bi composition. In Figure 5 we plot the electron-hole spatial overlap as a function of Al concentration in the barriers showing a strong improvement with overlap (and hence optical matrix element) with increasing barrier Al fraction. Increasing the Al fraction in the barrier reduces the refractive index contrast in the waveguide leading to a reduction in the optical confinement factor. Hence there is a compromise in terms of the impact of barrier Al fraction on the modal gain [35]. Note that the cladding Al fraction was fixed at 40% as a higher Al fraction would give rise to direct-indirect band gap cross-over in the cladding and additional processing issues associated with the higher propensity to oxidize.

Fig. 5. Dependencies of the spatial overlap of electron and heavy hole wave functions and optical confinement factor as a function of barrier Al fraction for a 7 nm wide GaAsBi/AlGaAs quantum well with 2.2% Bi at $T=295 \text{ K}$.

Based on preliminary calculations of the optical confinement factor of AlGaAs barriers as shown in Fig. 5, several laser structures were grown with 0%, 12% and 20% of Al the barriers/waveguides and different Bi composition in the QW(s). The QW width ($\sim 6.4 \text{ nm}$) was chosen to minimize the effects of interface inhomogeneity while maintaining suitable sub-band splitting. The structural details of the active region of the investigated laser diodes, calculated band offsets in the conduction (ΔE_C) and valence (ΔE_V) bands and transition energy (E_{E1-HH1}) at room temperature (RT) are summarised in Table I. The experimentally measured lasing wavelength (λ_{las}) and threshold current density (J_{th}) for these devices are also given in the table.

TABLE I
ACTIVE REGION PARAMETERS AND CHARACTERISTICS OF DIFFERENT
GAASBI LASER DESIGNS

	A	B	C	D	E	F
Number of QWs	1	3	1	1	1	3
QW width, nm	6.4	6.4	6.4	6.4	8	8
Bi%	2.2	2.2	2.2	2.2	4.4	~ 6
Al% in AlGaAs barriers	20	20	12	0	0	0
E_{E1-HH1} , eV	1.330	1.330	1.322	1.298	1.181	1.061
ΔE_C , meV	202	202	133	37	73	154
ΔE_V , meV	225	225	184	123	206	228
λ_{las} , nm	947	958	947	938	1039 ^a	1060
RT J_{th} , kA/cm ²	1.5-1.6	2.4-2.7	1.0-1.1	7.5	4.5 ^a	25

Table presents calculated band offsets in the conduction (ΔE_C) and valence (ΔE_V) bands, transition energy (E_{E1-HH1}) at RT as well as experimentally measured lasing wavelength (λ_{las}) and threshold current density (J_{th}) for these devices. Samples A, B, C, D and E were grown using MOVPE and for more details see [35], whereas the sample F was grown by the hybrid MBE/MOVPE method [46].

^a These data are measured at maximum operating temperature for this device at 180 K.

Figure 6 presents light output versus current density characteristics and stimulated emission spectra above lasing threshold for a set of four laser devices (A, B, C, and D) with nominally the same GaAs_{0.978}Bi_{0.022} single QW, with the exception of sample B which is a triple QW variant on structure A, and for different amount of Al in the barrier/waveguide layers. The J_{th} values in these devices at 295 K are presented in Fig. 2 with open circles [35]. It can be seen from Fig. 6 (a) and the data in Table I that there is a significant difference in the performance of these device with the lowest J_{th} of $\sim 1 \text{ kA/cm}^2$ being measured in the SQW laser with 12% Al in the waveguide (sample C). Figure 6 (b) also provides a comparison of stimulated emission spectra from these devices at current densities of $\sim 10\%$ above threshold at 295 K. Lasing operation at room temperature (RT) was observed in all of the devices containing 2.2% Bi with a wide range of threshold current densities as given in Table I. The conduction band offset, $\Delta E_C \approx 37 \text{ meV}$ in sample D with GaAs

barriers (see Table I) was relatively small and close to the thermal energy of carriers at RT. A strong thermal carrier spill-over in this structure resulted in a high threshold current density of 7.5 kA/cm² and, in addition to the lasing peak from the QW at ~938 nm, for temperatures >250 K, an emission peak from the barrier layers around 897 nm was observed as shown in Fig. 6 (b). The relatively broad lasing spectrum in this device at RT consisted of a multitude of Fabry-Perot modes due to a significantly broadened gain spectrum.

As can be seen from Fig. 6, there is an optimum Al concentration in the barriers to minimize the threshold current density for the 2.2% Bi containing QWs. To optimise the device performance, one should take into account two opposite effects, i.e. increasing band offsets for improved carrier confinement and reduction of the optical confinement factor and corresponding decrease in modal gain, related to the composition of Al in the waveguide and barrier layers, as per the data in Fig. 5.

Fig. 6. (a) Light-current characteristics and (b) lasing spectra at $T=295$ K of 2.2% Bi 6.4 nm SQW and TQW lasers with different Al composition in the waveguide/barrier layers (see details in Table I for the samples A, B, C, D).

Table I quantifies the influence of barrier Al fraction on the confining energy for electrons while Fig. 5 shows the effect of the waveguide Al fraction (all the structures here have Al_{0.4}Ga_{0.6}As cladding layers) on the optical confinement factor. Nominally identical laser structures A, C and D have an Al fraction in the barrier/waveguide layers of 20%, 12% and 0%, respectively. From Table I, it can be seen that the incorporation of Al in the barrier increases the electron confining potential from 37 meV for GaAs (0% Al) to

133 meV for 12% Al and 202 meV for 20% Al. For GaAs barriers, the confining potential is $\sim k_b T$ at room temperature (where k_b is the Boltzmann constant) whereas for both the 12% and 20% Al-containing barriers, the confining energy is $>5k_b T$ at room temperature, providing efficient electrical confinement of the carriers. Fig. 5 shows that the optical confinement factor changes from 1.21% to 0.89% for 12% and 20% Al-containing barrier/waveguide, respectively. Thus, in a simple approximation, one would expect the 12% Al barrier/waveguide structure (sample C) to provide >35% improvement in modal gain compared with the 20% Al barrier/waveguide structure (Sample A). The effect of this on device performance is clearly evident from Fig. 6 (a) and Table I, where we find that the sample C has a lower RT J_{th} of 1.0-1.1 kA/cm² compared with a substantially higher RT J_{th} of 1.5-1.6 kA/cm² in sample A. Thus, in spite of the better electrical confinement of carriers in the SQW device with 20% Al in the waveguide (sample A) compared to the corresponding 12% Al sample (sample C), the better optical confinement in the latter device resulted in an approximately 50% decrease in the RT J_{th} . On the other hand, as evidenced from sample D with a GaAs barrier/waveguide, the higher optical confinement factor cannot compensate for the very low electron confinement. These results demonstrate the need to carefully design the barrier/waveguide structure in QW lasers with low Bi content.

A comprehensive theoretical study of electrical and optical confinement in GaAsBi/(Al)GaAs laser structures was presented by Broderick et al. [50]. Their calculations suggest that QWs having barrier Al compositions 10-15% offer improved modal gain at fixed carrier density. Barriers with Al compositions less than 10% provide insufficient electron confinement to produce good levels of material gain, while for greater than 15% Al, the refractive index contrast between the barrier and cladding layers is reduced, eroding the optical confinement and hence the modal gain. According to their calculations, for Bi compositions 6% and more, GaAsBi/GaAs conduction band offset is sufficiently large that incorporating Al in the barrier layers has a negligible impact on the material gain and Al incorporation is not required in the barrier/waveguide layers.

Compared with the low Bi structures considered in Fig. 6, achieving the ultimate goal of Auger recombination and IVBA suppression requires substantially higher Bi fractions in the QW (~13% Bi for 1.55 μ m RT lasing wavelength [50]). The highest Bi composition realised up to now by MOVPE growth of 4.4% was achieved in the sample E, the details of which are provided in Table I. The threshold current density of these devices was substantially higher than in lower Bi-containing devices (samples A, B, C) resulting in a reduced maximum operating temperature of 180 K. In this device GaAs barriers were used, which may not be an optimum option for the trade-off between the electrical and optical confinement [50]. However, a comparison of light output versus current density characteristics of this structure with a nominally identical structure with lower Bi fraction of 2.2% (sample D), presented in Fig. 7 at $T=20$ K, shows that the threshold current density

for the sample E was as low as 1.13 kA/cm^2 compared with 1.36 kA/cm^2 for the sample D. Thus, in spite of the higher Bi content, at low temperature the 4.4% Bi devices E have a lower threshold current density than the 2.2% Bi D devices. The higher Bi fraction leads to larger band offsets and improved carrier confinement which, under the relatively high injection levels in SQW devices, offsets the material quality issues associated with the higher Bi fraction. However, the fact that the higher Bi fraction devices have poorer high temperature characteristics suggests that the defect-related recombination may be more important at higher temperatures, which needs the development of improved growth at higher Bi fractions.

Fig. 7. Light-current characteristics at $T=20 \text{ K}$ of 2.2% Bi and 4.4% Bi SQW structures which both have GaAs barriers/waveguides.

Fig. 2 shows that, in general, an increasing Bi fraction gives rise to an increase in J_{th} . Fig. 8 shows a summary of the temperature dependencies of J_{th} from some of the GaAsBi devices listed in Table I containing different Bi fractions and shows the variation in J_{th} with temperature and Bi fraction. This is a clear indication of the presence of loss processes which limit the device efficiency and performance at higher Bi fractions, as discussed in detail in section V of this paper.

Fig. 8. Temperature dependence of threshold current density in several GaAsBi/(Al)GaAs devices with different Bi composition as shown in Table I.

In the structure with the highest Bi fraction of $\sim 6\%$ (sample F), J_{th} was very high ($\sim 25 \text{ kA/cm}^2$) at RT, which is 25

times greater than its value in the best single QW devices with 2.2% Bi (sample C). Interestingly, despite a significant change in absolute values of J_{th} in the different samples in Fig. 8, the relative change of J_{th} with temperature, particularly in the region of interest around RT, characterised by T_0 is very similar in various bismide devices reported up to now, where the T_0 value is around 90-110 K [35, 43, 46]. Taking into account the relatively high J_{th} even in low Bi containing samples this indicates that at the devices are dominated by a similar dominant recombination or loss process.

We should also note here that measurements of pure spontaneous emission spectra in bismide lasers (discussed in section V) indicate a high degree of inhomogeneous broadening of carrier distribution function due to inhomogeneity of the QW(s) width as well as non-uniform Bi composition in the active region [35, 46, 51]. The inhomogeneous carrier distribution causes localisation effects at low temperature and improved carrier thermalisation with increasing temperature, which is responsible for the regions of unusual temperature insensitive J_{th} or regions with negative T_0 as can be seen for sample F in Fig. 8 [46]. It was earlier reported that inhomogeneous carrier distribution in sample F coupled with non-radiative recombination caused a very high RT threshold current density and a strong blue shift of the facet emission peak of $\sim 100 \text{ meV}$ with increasing current density from 0.1 kA/cm^2 up to 20 kA/cm^2 [46]. Such a strong blue shift of the lasing peak relative to the E_{E1-HH1} transition energy almost entirely compensated the composition-induced reduction in E_g achieved in structure F compared to structure E (see Table I).

Inhomogeneity often causes non-pinning of the carrier density in these devices above threshold, which in turn, in the presence of non-radiative losses, can further degrade the laser performance by reducing the laser light output and the slope efficiency [35]. The observed laser characteristics are presently inferior to commercial near-infrared lasers operating around $1 \mu\text{m}$. However, this reflects the early stage of development of GaAsBi based lasers, where the focus has been on material development to push laser operation towards telecoms wavelengths. Further laser structure design and optimization are underway for improvement of the device performance.

V. RECOMBINATION MECHANISMS

To study recombination processes in bismide lasers, particularly in order to understand the origin of the relatively high J_{th} values, two main experimental techniques were used. Using direct measurements of pure spontaneous emission through a $100 \mu\text{m}$ in diameter circular window in the substrate contact [3, 56], the temperature dependence of the radiative component (J_{rad}) of the threshold current density was analysed [35]. In this method, J_{rad} is determined as the integrated spontaneous emission at threshold, which is proportional to the radiative recombination rate. By measuring this as a function of temperature, both J_{th} and J_{rad} can be measured independently. By assuming negligible losses at the lowest temperature (here it is 50 K) or, in other words, $J_{th} = J_{rad}$, an

estimation of the *upper limit* for J_{rad} and the *lower limit* for the non-radiative component (J_{nonrad}) of J_{th} at higher temperatures is possible. As an example, Fig. 9 shows that even in the best laser structure (sample C, see Table I) the main contribution to J_{th} is due to a non-radiative recombination process. Despite the difference in absolute values, both of the 2.2% Bi SQW structures A and C showed a very similar temperature dependence of J_{th} with a characteristic temperature of ~ 130 K below 300 K and $T_0 \sim 100$ K at 300-350 K. At $T > 200$ K, an anomalous decrease of J_{rad} was observed because of the increasing absorption of the spontaneous emission by the GaAs substrate [35, 51]. To take into account this effect, since $J_{rad} \propto T$ in an ideal QW laser [3], a linear interpolation of J_{rad} at low temperatures in Fig. 9 was used to estimate J_{rad} at RT. Such a method of correction was supported by segmented contact method which showed that the real value of integrated spontaneous emission is $\sim 30\%$ higher compared with the one measured from the substrate window, details of which are published elsewhere [51]. Taking this into account, we found that J_{rad} at RT accounts for up to 20% of J_{th} at room temperature. From Fig. 9 one can see that J_{nonrad} , which is the difference between J_{th} and J_{rad} , makes up at least 80% of the threshold current at RT. This is a significant amount of loss, considering that at these wavelengths ($< 1 \mu\text{m}$) and with high enough band offsets in these structures, the other loss processes such as Auger recombination (typically important in lasers emitting above $1.3 \mu\text{m}$ [3, 59]) or carrier leakage are both expected to be small.

Fig. 9. Temperature dependence of J_{th} and its radiative (J_{rad}) and non-radiative (J_{nonrad}) components in sample C (see Table I). J_{rad} was normalised to J_{th} at 50 K, assuming $J_{nonrad} = 0$, to estimate maximum value of ratio J_{rad}/J_{th} and minimum fraction of “non-radiative” current (J_{nonrad}/J_{th}) at RT. The solid curves represent a guide to the eye, whereas the dashed curves are corrected dependencies to take into account absorption of spontaneous emission in the substrate at increasing T .

To investigate the recombination and loss processes in these devices in more detail we carried out measurements of J_{th} and J_{rad} as a function of high hydrostatic pressure. The application of pressure to a semiconductor device provides a useful means of reversibly varying the band gap at constant temperature in a fully-functioning device [56, 59]. The radiative and various

non-radiative recombination processes have characteristic dependencies on the band gap, and hence high pressure can be used to identify the dominant recombination mechanisms [56, 59, 60]. From Fig. 9 it is apparent that even in the best 2.2% Bi QW devices reported here, more than 80% of J_{th} occurs through non-radiative recombination. Considering the pressure dependencies of different mechanisms, Auger recombination decreases strongly with increasing pressure, which is increasing E_g , and, hence, in Auger dominated lasers, J_{th} decreases with increasing pressure [56, 59]. Carrier leakage into indirect satellite valleys, as often observed in visible lasers and quantum cascade lasers [57, 61] occurs when the energy separations between direct and indirect minima of the conduction band (Γ -X or Γ -L) are small. Such a process typically gives rise to an exponential increase in J_{th} with pressure due to the fact that pressure causes a decrease in the Γ -X and Γ -L separation [56, 59]. However, in the GaAsBi devices studied here, the low Al composition in the barriers causes a large separation Γ -X or Γ -L. Therefore, indirect inter-valley carrier leakage is also not expected to be important in these devices. Non-radiative Shockley-Read-Hall (SRH) recombination due to defects occurs due to carriers recombining via localised states, releasing their energy in the form of phonons. This process and the associated current is usually independent of pressure because shallow defect levels stay closely tied to the conduction band edge as hydrostatic pressure is applied [62]. In an ideal QW laser, the radiative part of threshold current increases approximately as E_g^2 [56, 59, 60] and for such devices, a similar weak increase in J_{th} with pressure is expected.

A detailed description of the experimental procedure and more discussion of the pressure data can be found in [35, 46, 59]. Here we mainly consider the pressure dependencies of J_{th} and J_{rad} with the aim to identify the dominant loss mechanism. To measure J_{rad} under pressure, we processed sample A into circular mesa devices with the diameter of 50 μm , 100 μm and 200 μm as shown in the inserted photo in Fig. 10 (a). The devices were cleaved and wire bonded onto standard transistor-outline headers. Then, integrated spontaneous emission was measured as a function of injected current at different pressure values. Having previously measured the pressure dependence of J_{th} in laser devices [35], J_{rad} was determined from mesa devices as the integrated spontaneous emission value at the corresponding J_{th} value for each pressure. Figure 10 (a) presents the pressure dependence of J_{rad} (open circles) normalised to its value J_{rad0} at ambient conditions. To estimate pressure dependence of E_g we used the pressure dependence of the photon energy of lasing emission (E_{las}) [35]. In Fig. 10 (a) we also plot the normalised pressure dependence of J_{rad} in an ideal QW as E_{las}^2 , which appeared to show a smaller increase with the hydrostatic pressure compared to J_{rad} in sample A. In this case, J_{rad} can be fitted with a simple empirical power law function having a stronger power dependence on the bandgap, where we find that a curve for which $J_{rad} \propto E_{las}^{3.6}$ provides a good fit (see solid line in Fig. 10 (a)). In our previous pressure studies of self-assembled InAs/GaAs quantum dot (QD) lasers we observed an even

stronger increase of J_{rad} with increasing pressure, where $J_{rad} \propto E_{las}^6$ [63]. A characteristic feature of self-assembled QDs is a significant inhomogeneous broadening of the carrier distribution functions due to variation in QD sizes [59, 63], which causes broadening of the gain spectrum. The variation in J_{rad} with pressure observed here for the bismide QW devices suggests that similar reasons are responsible for the stronger bandgap dependence of J_{rad} compared to an ideal QW laser.

The pressure dependence of J_{th} for sample A (NB: a similar dependence was also observed in samples B, C) is shown in Fig. 10 (b). From the earlier spontaneous emission analysis (see Fig. 9) we may estimate that $J_{rad}/J_{th}=0.2$ at RT and atmospheric pressure and using the measured pressure dependence of J_{rad} from Fig. 10 (a) it was possible to determine J_{nonrad} ($J_{nonrad}=J_{th}-J_{rad}$) and its relative change with increasing pressure, which is plotted in Fig. 10 (b). One can see from Fig. 10 that J_{nonrad} is almost independent of pressure, which is consistent with defect-related recombination dominating the non-radiative path.

Fig. 10. (a) Normalized high hydrostatic pressure dependence of the radiative part (J_{rad}) of the threshold current measured using sample A and specially fabricated circular-mesa devices as shown in the inserted photo. The solid line is a fit of the data using power dependence of bandgap (E_{las}). The dashed line is the dependence E_{las}^2 , which is corresponding to the pressure dependence of J_{rad} in an ideal QW laser [60]; (b) The normalized threshold current density variation with pressure and its radiative and non-radiative components determined using pressure data in Fig. (a) and assuming $J_{rad}=0.2J_{th}$ from spontaneous emission measurements as shown in Fig. 9. J_0 are the corresponding current densities of different components and total current density at atmospheric pressure.

The pressure dependence of the normalised threshold current density of a 1 mm long and 50 μm wide Fabry-Perot laser with the highest Bi fraction of $\sim 6\%$ (sample F) at $T=76$ K is presented in Fig. 11. The measurements at a low temperature of 76 K minimise carrier leakage (since

$kT \approx 6$ meV, much smaller than the band offsets) as well as Auger recombination which is negligible at these wavelengths at low temperature. The observed almost constant pressure dependence of J_{th} indicates that the dominant recombination process occurs via defects even at low temperature as observed in the 2.2% Bi devices at room temperature (see Fig. 10 and [35]). Defect-related (SRH) recombination therefore appears to dominate device performance giving rise to the high threshold current densities measured in these initial devices. Such a process would also help to explain the relatively constant T_0 value of ~ 100 K around RT as reported by several groups in different devices with different Bi compositions [35, 43, 44].

Thus, from the temperature and pressure dependencies of J_{th} and J_{rad} it was possible to identify that non-radiative current path is dominated by SRH defect related recombination, which accounts for more than 80% of J_{th} in the 2.2% Bi QW lasers and is responsible for the very high J_{th} in the devices with higher Bi composition. We, therefore, conclude that defect-related recombination dominates the devices and that further optimisation of the growth of the GaAsBi/AlGaAs laser structures is necessary to improve the performance of the devices, setting a particular challenge for achieving 1.55 μm (Bi $\sim 13\%$) Auger-free operation on GaAs. We note here that defect-related recombination was similarly a problem for dilute nitride based lasers in the 1.3-1.6 μm range [64, 65, 66]. However, substantial efforts from a large number of groups in growth optimisation led to significant improvements in material quality with consequent improvement in device performance [67, 68, 69]. The bismides are somewhat less mature and at a much earlier stage of material optimisation. Importantly, however, we note that while the dilute nitrides continued to suffer from Auger recombination, optimised bismide based device potentially offer a unique route to fundamentally suppress this loss process.

Fig. 11. High pressure dependence of the threshold current density (J_{th}) of the 6% Bi TQW device (sample F in Table I) normalised to its value at atmospheric pressure (J_{th0}) measured at low temperature of 76 K.

VI. FUTURE POSSIBILITIES: GAASBI(N) AND INGAASBI BASED LASER STRUCTURES

In this section we briefly consider several other opportunities which are opened-up by using bismide III-V alloys in various photonics applications.

A. GaAsBi(N) based type-I and type-II heterostructures

Adding a small amount of nitrogen to the GaAsBi alloy system adds a wide range of possibilities to provide flexible control of the band gap, band offsets, strain and spin-orbit splitting of the GaAsBiN material system as shown in Fig. 12 from which it can be seen that the preferential band structure for CHSH Auger and IVBA suppression ($\Delta_{SO} > E_g$) can be obtained in the mid-infrared on a GaAs system [17, 70]. Furthermore, due to the specific nature of N and Bi atoms to interact mainly with the conduction and valence band, respectively, it is also possible to develop not only a wide range of type-I hetero-structures [17, 70], but also strain balanced GaAsBi/GaAsN type-II heterostructures which can help to span the operating wavelength further into mid-infrared [71]. These novel systems demonstrate the significant potential of this new class of III-V heterostructures for mid-infrared applications on GaAs.

Fig. 12. Predicted band gap and spin-orbit splitting energy as a function of Bi and N compositions in GaAsBiN/GaAs. Insert demonstrates a possibility to control strain by adjusting Bi and N compositions.

B. InGaAsBi/InP structures for mid-infrared applications

It has been shown that the InGaAsBi material system grown on conventional InP substrates is also very promising for near- and mid-infrared photonic devices operating in the spectral range of 1.5–4.1 μm [28, 29]. Similarly to GaAsBi, in this system it is possible to realize the condition $\Delta_{SO} > E_g$, but at significantly lower Bi fractions (>3-4%) as presented in Fig. 13 [28, 29]. Consequently, this has the potential to improve the high-temperature performance and thermal stability of mid-infrared photonic devices with the distinct advantage of being able to produce them on InP using standard telecoms laser fabrication processes. The detailed discussion of the effect of Bi on the bandgap, spin-orbit splitting energy, band offsets and strain of InGaAsBi are discussed in further detail in [29, 59].

Fig. 13. Predicted E_g of InGaAsBi/InP and experimental data at 300 K demonstrating a wide scope for band structure engineering in the range of 0.3–0.8 eV (1.5–4 μm) with $\Delta_{SO} > E_g$ at Bi fractions >3-4%.

VII. CONCLUSIONS

In this paper, we have reviewed the potential of bismide III-V alloys for applications in near- and mid-infrared photonic devices owing to the unique properties of Bi to manipulate the valence band thereby providing strong control of the bandgap and spin-orbit splitting energy. It has been demonstrated that Bi-III-V alloys can offer the potential to obtain materials with bandgaps covering the near- and mid-infrared ranges whilst also providing the beneficial condition that $\Delta_{SO} > E_g$, which would allow for the suppression of the major Auger recombination and optical loss processes which plague lasers and LEDs at these wavelengths. Using quaternary systems like GaAsBiN on GaAs substrates or InGaAsBi on conventional InP substrates provides even more flexibility in controlling both strain and conduction and valence band properties, giving the possibility to develop various type-I and type-II hetero-structures.

We have reviewed current achievements in growth and bismide device fabrication and presented some experimental and theoretical results on existing state-of-the-art GaAsBi lasers. Our results illustrate that while present devices are limited by defect-related recombination, if this can be minimized there is wide-ranging potential for the bismide III-V material system for the development of efficient and temperature stable photonic devices covering a wide spectral range with applications in the near- and mid-infrared.

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